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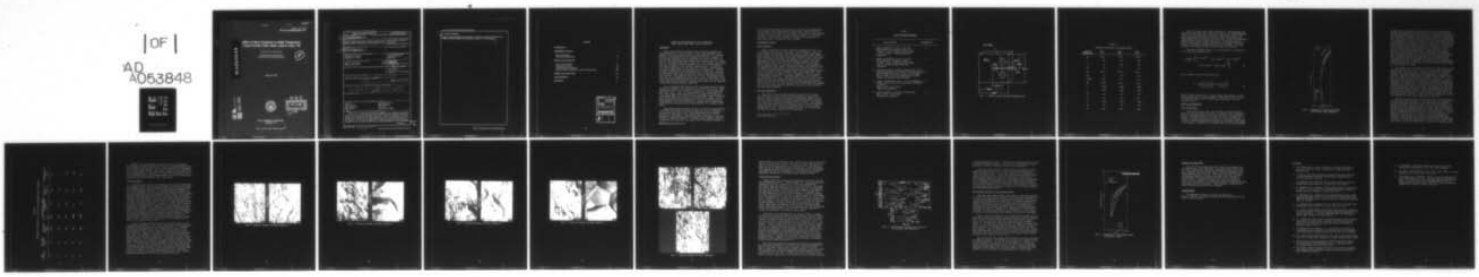

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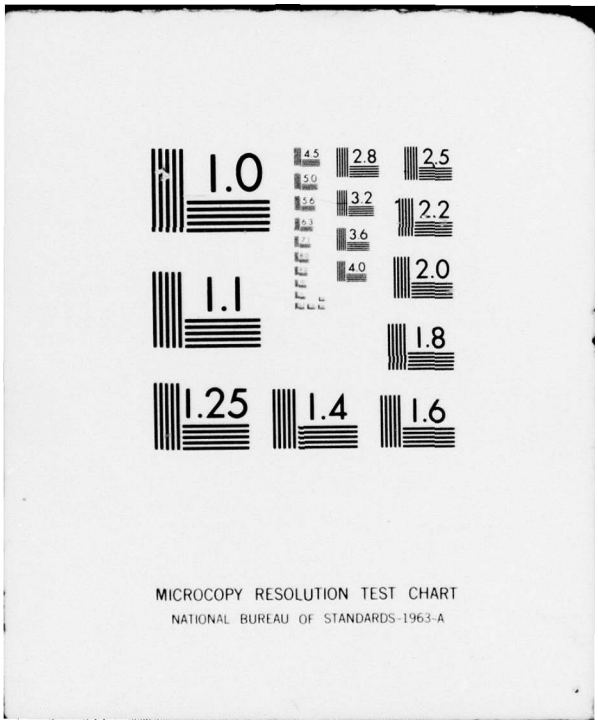
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Effect of Heat Treatment on High Temperature Crack Growth Under Static Load in Alloy 718

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Thermostructural Materials Branch
Material Science and Technology Division

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Heat treatments	High temperature
Creep crack growth	Static load
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Alloy 718 in its conventional heat treated form possesses very poor resistance to crack growth under static load in the temperature range of 425 to 760°C. Since this crack growth is found to be sensitive to the microstructure, a detailed investigation of the effect of heat treatment on the crack growth behavior has been made and various heat treatments have been evaluated with respect to their resistance to this crack growth. Among the heat treatments evaluated it is shown that solution at 926°C for 10 hours and age at 734°C for 48 hours would significantly improve the crack growth

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20. Abstract (Continued)

resistance without sacrificing its tensile properties. This improved resistance is attributed to the formation of Ni_3Cb precipitates which effectively strengthen the grain boundaries.

CONTENTS

EFFECT OF HEAT TREATMENT ON HIGH TEMPERATURE CRACK GROWTH UNDER STATIC LOAD IN ALLOY 718

BACKGROUND

Alloy 718, because of its high tensile and creep strengths and good weldability, is used extensively in many high temperature components, such as gas turbines, and is being considered as a candidate material for applications in fast breeder reactors (1). However, it has been shown recently that at temperatures of 425°C and above, the alloy in its conventional heat treated form possesses very poor resistance to crack growth under static load (2-4). For example, crack growth can occur at 650°C at stress intensities as low as 15 MPa$\sqrt{m}$ which is less than 15 percent of its fracture toughness value. In addition, once crack growth starts at the surface of a compact tension specimen, it grows so rapidly that the specimen fails in a few hours. In fact, the stress intensities and crack growth rates at 650°C are comparable to those under cyclic load (5) when the rates are expressed on a time basis (da/dt). This implies that static loads at this temperature for this alloy are as damaging as cyclic loads. Crack growth under static load at such low stress intensities also implies that the critical size cracks that can be tolerated in a structure are small and the components of this alloy require frequent inspection.

One of the significant aspects of the crack growth behavior under static load, however, is its sensitivity to microstructure. For example, overaging of Alloy 718 increases its resistance to crack growth under static load (2,3). Also in a conventionally heat treated Udimet 700 which has a large volume percentage of precipitated hardening phase, crack growth under static load occurs at much higher stress intensities and at much lower rates than that in Alloy 718 (6). Wilson (7) has shown recently that notch sensitivity of several nickel-base alloys, including Alloy 718, can be improved by changing heat treatment which could result in a larger size and volume fraction of the precipitated phase.

Conventional heat treatments for Alloy 718 have been developed (8) to optimize its tensile and creep strengths but not its resistance to crack growth under static load. Since conventional heat treatments provide poor resistance to crack growth under static load, it is important at this stage to evaluate other heat treatments to determine if they can improve resistance. A desirable heat treatment could be selected for the alloy that could give rise to increased resistance to

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this crack growth without sacrificing its tensile and creep strengths. In this report crack growth behavior under static load at 650°C for five different heat treatments is presented and the results are analyzed in terms of linear elastic fracture mechanics concepts. In addition, the effect of heat-to-heat variation on crack growth resistance is determined using a conventional heat treatment.

EXPERIMENTAL DETAILS

Heat Treatments

Table 1 lists the five heat treatments designated as A, B, C, D, and E that were selected for this study. Heat treatments A and B are conventional heat treatments, the first one with a lower temperature anneal, 955°C, and the other with a higher temperature anneal, 1050°C. Heat treatment A is the same as that used previously (3,4). Heat treatment C was developed by Muller and Donachie (9) to improve the notch sensitivity of the alloy. This heat treatment involves a two-step anneal with 20 min at 1035°C and 10 hr at 915°C before it is aged, with the aging similar to that in heat treatment A. Heat treatment D was developed by Idaho National Engineering Laboratory* (INEL) in order to minimize the susceptibility of the alloy to weld cracking (10). This heat treatment involves a high temperature solution anneal (1093°C for 1 hr) and slow cooling (38°C/hr) to the aging temperature. The aging procedure is the same as in heat treatment A. Heat treatment E was developed by Wilson (7). Annealing temperature is the lowest of the above (926°C for 10 hr). Aging is done at 734°C for 48 hr. Wilson argues that this heat treatment results in larger size precipitates which could be responsible for the improved time dependent notch sensitivity of the alloy. We shall discuss in more detail later the probable microstructures resulting from each heat treatment and their effect on the crack growth behavior.

Specimen Preparation

The Alloy 718 used for this study was in the form of 15.3-mm thick plate from a vacuum induction melted-vacuum arc remelted heat. Modified compact tension (CT) specimens of dimensions shown in Fig. 1 were cut from the plate with the notch along T-L orientation and then were given one of the heat treatments listed in Table 1. The composition of the heat supplied by the vendor is listed in Table 2 under the designation V18 heat along with the compositions of the other heats tested earlier (3-5). Tensile specimens were also cut from the plate and were heat treated along with the CT specimens.

*Now EG&G Idaho, Inc.

Table 1

Heat Treatments Employed

Heat Treatment	Designation
1. Anneal at 955°C for 1 hour - air cool Age at 720°C for 8 hours, furnace cool to 620°C. Hold at 625°C for total aging time of 18 hours. (Conventional heat treatment).	A
2. Anneal at 1050°C for 1 hour - air cool Age at 760°C for 10 hours, furnace to 650°C. Hold at 650°C for total aging time of 20 hours (Conventional heat treatment).	B
3. Anneal at 1035°C for 20 min, furnace cool to 915° Intermediate anneal at 915°C for 10 hours, air cool Age at 720°C for 8 hours, furnace cool to 620°C. Hold at 620°C for total aging time of 18 hours. (Muller and Donachie heat treatment)	C
4. Anneal at 1093°C for 1 hour - cool at 38°C/hour to 720°C - Age at 720°C for 4 hours cool at 38°/hour to 620°C - Hold at 620°C for 16 hours (INEL heat treatment).	D
5. Anneal at 926°C for 10 hours, furnace cool Age at 734°C for 48 hours (Wilson heat treatment)	E

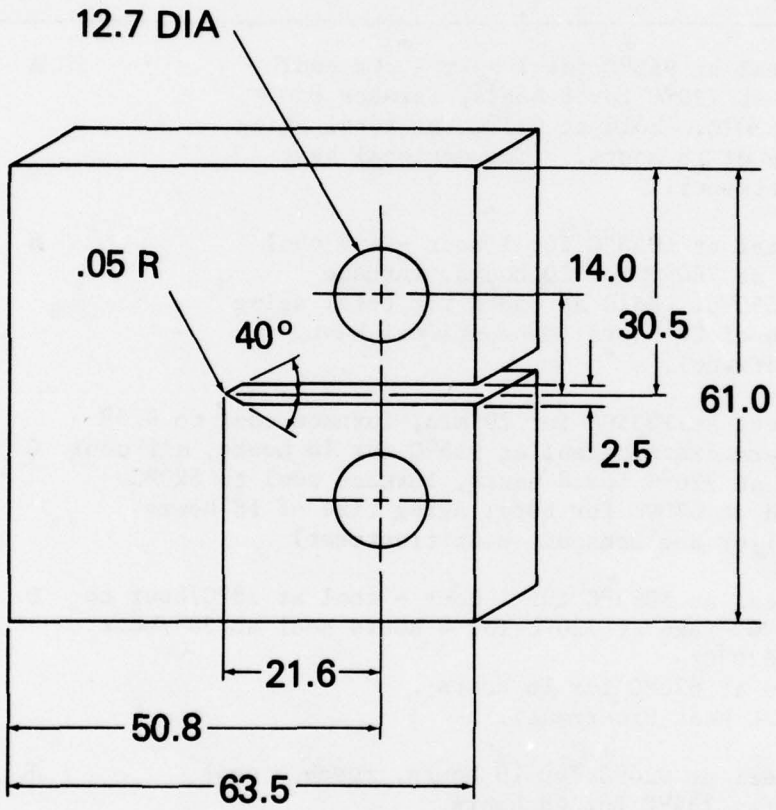

Fig. 1 - Compact tension specimen dimensions (mm)

Table 2

Chemical Composition of Different Heats

Chemical Composition	Heat V17	Heat V18	Heat V47
Cr	18.8	18.15	17.96
C	.05	.05	.04
Si	.17	.08	.05
Co	.25	.4	.43
Ni	51.07	51.52	51.14
Mn	.13	.10	.19
Cb+Ta	5.06	5.14	5.19
Mo	3.09	3.13	3.12
P	<.005	<.005	<.005
S	<.005	<.005	<.005
Al	.5	.57	.52
Ti	.99	.99	1.0
B	.005	.004	.003
Cu	.02	.02	.02
Fe	Bal.	Bal.	Bal.

The CT specimens were fatigue precracked at room temperature and were loaded in a creep frame. They were heated to the test temperature, 650°C, by a resistance wire-wound furnace containing a window. The crack lengths were measured periodically using a low magnification travelling microscope and the crack growth rates, da/dt, were determined from the plots of crack length versus time. The tensile tests at 650°C were performed on a MTS machine and the extension was measured using a strain gauge extensometer afixed to the specimen. The load-extension curves were recorded directly using an X-Y recorder.

The stress intensity factor, K, was calculated for a/W values up to 0.7 from the expression (11),

$$K = \frac{P}{BW^{0.5}} \left(29.6\ (a/W)^{0.5} - 185.5\ (a/W)^{1.5} + 655.7\ (a/W)^{2.5} - 1017.0\ (a/W)^{3.5} + 638.9\ (a/W)^{4.5} \right) \quad (1)$$

and for a/W>0.7 from the relationship (12),

$$K = \frac{P}{2B} \left(\frac{W + a}{(W-a)^{1.5}} \right) \left(4.0 + \frac{(W - a)}{(W + a)} \right) \quad (2)$$

where P is the applied load, a is the average crack length, W is the specimen width, and B is the thickness. Correction for the curvature of the initial crack front was made using measurements from the fracture surfaces.

RESULTS AND DISCUSSION

Crack Growth Rates

Crack growth rates, da/dt, as a function of stress intensity factor, K, are represented in Figure 2 for different heat treatments. The heat treatments given and the loads applied are also shown in the figure. For heat treatment E, the data for only one specimen are reported. Significant tunneling occurred for this heat treatment and in fact the second specimen failed without any observable crack growth on the surface. Therefore, crack growth rates for this heat treatment

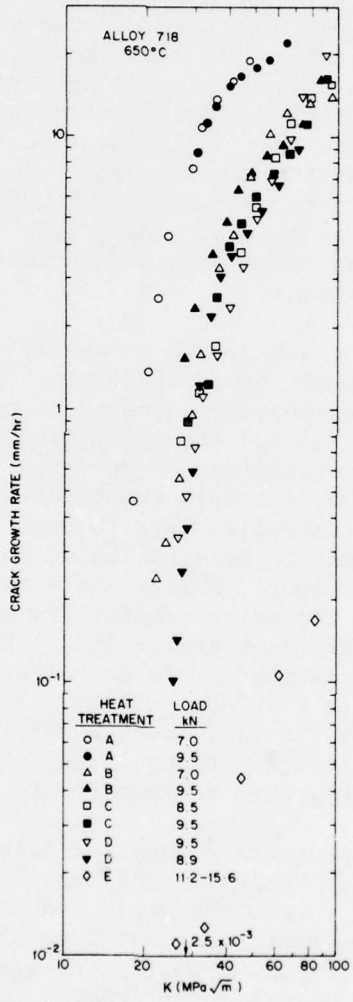

Fig. 2 - Comparison of crack growth rates for different heat treatments

in Figure 2 were obtained by intermittently fatigue precracking the specimen and measuring the crack length from the fracture surfaces. The two data points representing the lowest crack growth rates for this heat treatment correspond to very small crack growth at the mid-thickness region which occurred when the specimen was kept at a lower load for a long time. The two points therefore fall in the region close to a threshold stress intensity. There is a large disparity in the crack growth behavior of the specimens of heat treatment E in comparison to those of the other heat treated specimens. For example, in all other heat treatments crack growth occurred at the surface and it was continuous until the specimens failed. Most important to note in Figure 2 is that crack growth rates in specimens that underwent heat treatment E are nearly two orders of magnitude smaller than the growth rates in specimens that underwent the other heat treatments. Also of interest is that the lower temperature conventional heat treatment (A) results in the highest crack growth rates compared to those from other heat treatments. The growth rates for heat treatment A are nearly the same as those obtained earlier using different heats with the same heat treatment.

The resistance to crack growth is nearly the same for heat treatments B, C and D. They all have relatively high solution annealing temperatures, 1035°C and above. Among these three heat treatments, however, the cooling rates and the aging treatments are slightly different. The aging temperature is somewhat higher for heat treatment B, (760°C) than that for heat treatment C and D (720°C). Specimens were cooled at a controlled rate in heat treatment D. These differences do not appear to be significant in terms of crack growth resistance under static load. Miller and Donachie (9) developed heat treatment C to improve the notch sensitivity of Alloy 718. The crack growth resistance of this heat treatment, although greater than that of conventional heat treatment A, is not significantly different from those of heat treatments B and D. Wilson (7) has claimed that heat treatment E makes Alloy 718 less notch sensitive compared to the two conventional heat treatments. Figure 2 shows that this heat treatment also improves the crack growth resistance of this alloy.

Hardness values and tensile data for Alloy 718 for different heat treatments are listed in Table 3. The conventional heat treatment A provides the highest hardness and yield and tensile strengths but average ductility. The heat treatment B results in relatively lower hardness and yield strength but nearly the same tensile strength as the first one. The heat treatments C and D have much lower yield and tensile strengths than those of A and B. The ductility B, C and D are nearly the same but it is less than that of heat treatment A. Most important to note in Table 3, however, is that heat treatment E results in yield strength close to that of the conventional heat treatment A, but gives ductility greater than those of the above heat treatments. We may also mention here that grain size for heat treatment A and E are the same while for B, C and D it is slightly larger.

Table 3

Mechanical Properties of Alloy 718 for Different Heat Treatments

Heat Treatment	Room Temp. Hardness R_c	Properties at 650°C				
		.2% yield Stress MPa	U.T.S MPa	% Elongation at U.T.S.	Total Elongation	% Reduction in Area
A	46	1004	1158	9.5	18.7	15
B	43	964	1155	10.8	14.1	5
C	42	852	994	12.6	14.1	2.6
D	41	874	1090	11	14.3	5.6
E	44	948	1085	7.6	26.6	36

Comparison of the mechanical property data with crack growth data in Fig. 2 shows that neither the grain size, yield strength, tensile strength, nor the ductility has a unique correlation with respect to the crack growth behavior. Since all of the tests were conducted in air, the differences in the growth behavior cannot be attributed solely to environment either. Although all of the above factors may play a role in determining the crack growth behavior, none of them alone determines the crack growth behavior under static load.

Fracture Surfaces

In order to get a better understanding of the factors responsible for the differences in the crack growth behavior, the fracture surfaces were examined under a field emission scanning electron microscope and the results are discussed below. Crack growth was intergranular for all of the specimens. Details of the fracture surfaces, however, provide some understanding of the microstructural features of the specimens that have been heat treated differently. Figure 3 shows the surface features of the specimens of heat treatment A under low and high magnification. The grain size is relatively small and the grain surfaces are covered by second phase particles which are either inherent in the structure or formed by oxidation. Figures 4, 5, and 6 show the surface features of the specimens that underwent heat treatments B, C, and D, respectively. Figures 4 and 6 appear to be nearly the same with relatively clean surfaces. This implies that in terms of microstructural features on the grain boundaries, controlled cooling (heat treatment D) is not much different from the conventional heat treatment B. Among the three heat treatments, heat treatment C appears to result in slightly larger precipitates on the grain boundary surfaces as can be seen in Fig. 5. But this difference appears to be insignificant in terms of crack growth rates in Fig. 1.

The most striking difference in terms of microstructure is that given by heat treatment E. The fracture surfaces of the specimens that underwent this heat treatment are represented in Fig. 7. Comparison of previous fractographs with Fig. 7 shows that the amount of the second phase particles is much larger than that given by any other heat treatment. Figure 7b is obtained in the slow crack growth region while Fig. 7c is obtained in the fast fracture region with features characteristic of microvoid coalescence. A similarity in the fracture surfaces in Fig. 7b and 7c implies that in this heat treatment, crack growth under static load might have occurred by the nucleation of voids ahead of the main crack and their subsequent coalescence with the main crack. In the other heat treated specimens no such void coalescence is apparent from the fracture surfaces. Presence of significant quantity of second phase particles on the grain boundaries could be responsible for a change in the crack growth mechanism from a diffusion-controlled process to a deformation-controlled process (13). Similar reasoning also accounted for the large difference in crack

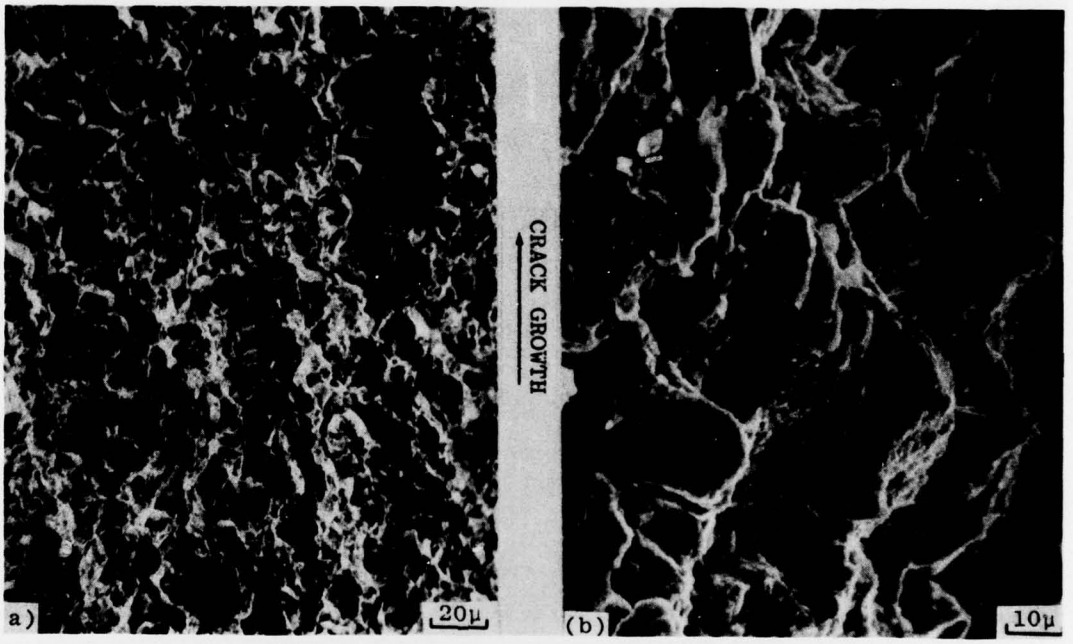

Fig. 3 - Fracture surfaces for heat treatment A

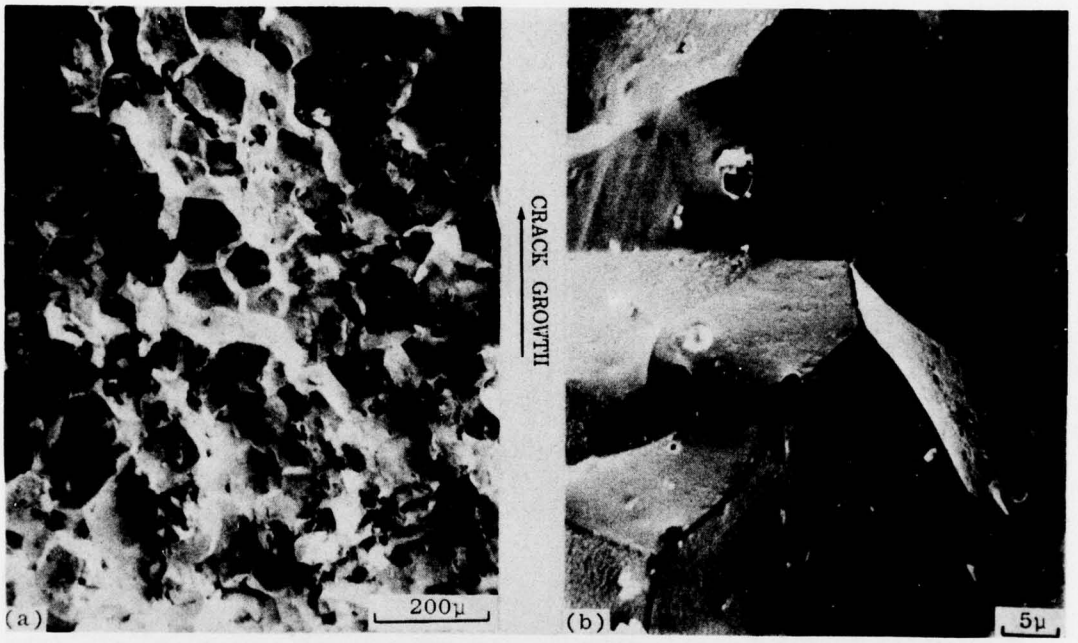

Fig. 4 - Fracture surfaces for heat treatment B

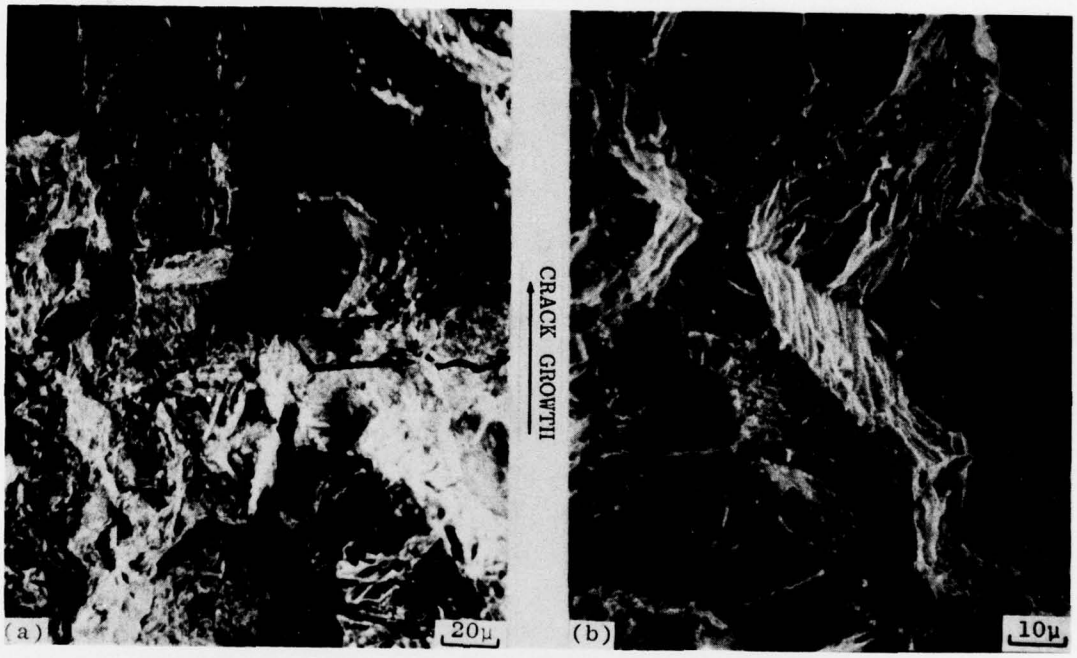

Fig. 5 - Fracture surfaces for heat treatment C

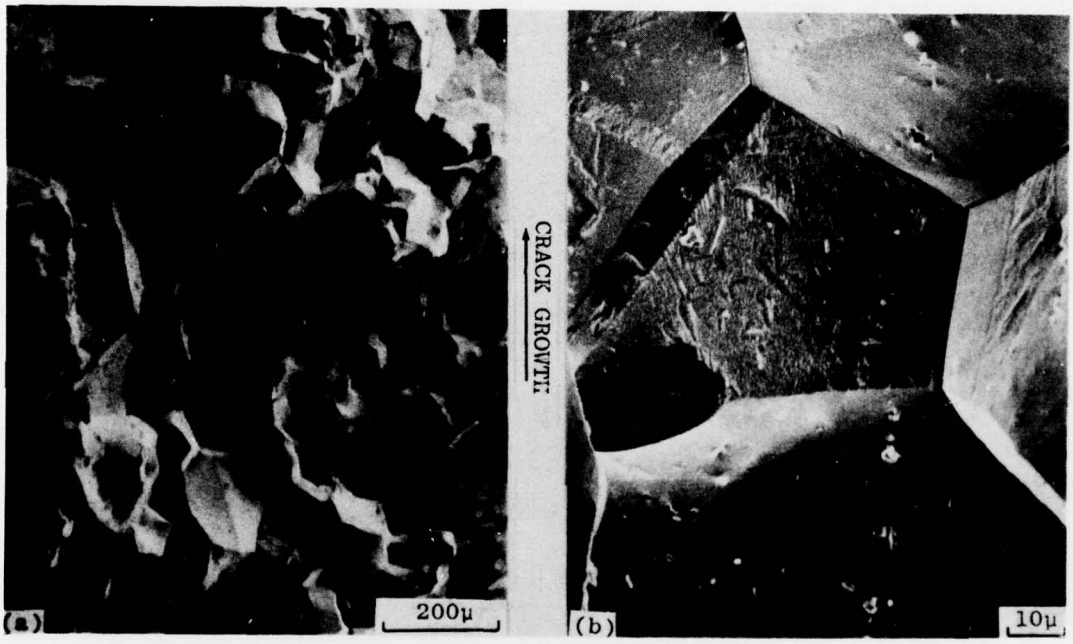

Fig. 6 - Fracture surfaces for heat treatment D

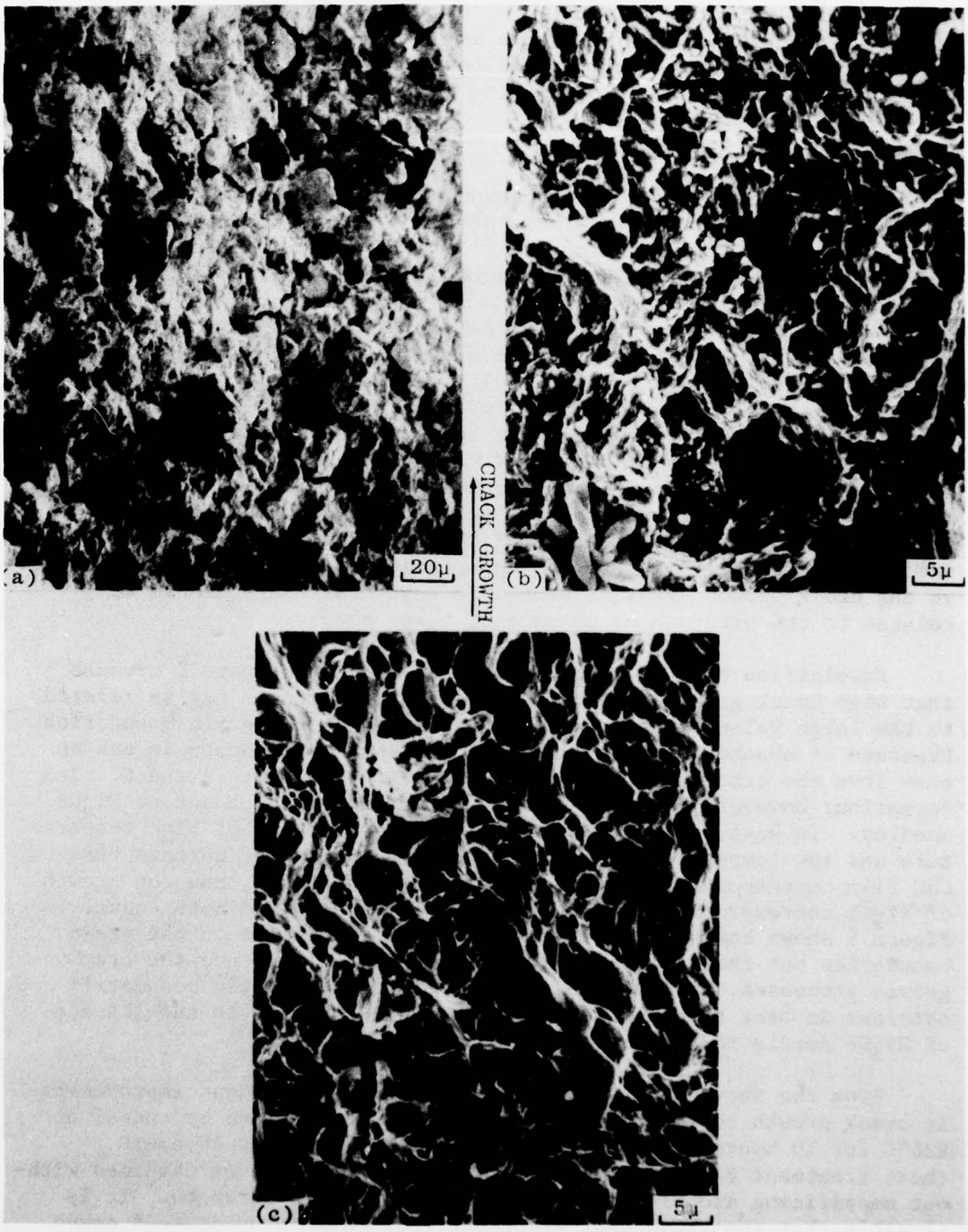

Fig. 7 - Fracture surfaces for heat treatment E

growth behavior in conventionally heat treated Udimet 700 and Alloy 718. The crack growth rates in Udimet 700, which contains relatively large amount of precipitated phase, are significantly lower than those in Alloy 718 in its conventionally heat treated form. Using transformation diagrams we shall next inquire as to the identity of the precipitate particles that are responsible for the improved crack growth resistance, and methods of optimizing their volume fractions.

Transformation Diagrams

Figure 8 shows the transformation diagrams for Alloy 718 for the heat treatments in Table 1. These diagrams were constructed using the information contained in Reference 8. There is a considerable uncertainty about the position of some of the boundaries. The reason for this is that the transformations in this alloy are very sluggish and also the formation of some of the phases depends on solution time and temperature as well as processing history. For example, high solution temperatures such as in heat treatments B and D and probably in C favor the formation of CbC film on grain boundaries. Such formation is diminished for low solution temperatures such as in A and E and this is related to the inability of the previously formed carbides to go into solution at the low temperatures (14). Likewise the formation of Ni_3Cb needles is favored by low solution temperatures and long times. Since only the formation of CbC film and the Ni_3Cb needles varies significantly between the heat treatments, the large differences in the crack growth behavior among the heat treatments should be related to the presence or absence of these phases.

Examination of crack growth data in light of Figure 8 reveals that high crack growth resistance for heat treatment E may be related to the large volume fraction of Ni_3Cb needles at the grain boundaries. Presence or absence of CbC film may not be of much concern as can be seen from the crack growth behavior for heat treatments A and B. Its formation, however, could deplete Cb and could limit amount of Ni_3Cb needles. In heat treatment C, there is a combination of high temperature and low temperature solution anneal and it is not certain that CbC film corresponding to high temperature anneal and enhanced growth of Ni_3Cb corresponding to low temperature anneal would both occur. Figure 5 shows however some formation of Ni_3Cb needles on the grain boundaries but their amount may be too small to influence the crack growth processes. Likewise the relatively cleaner grain boundaries obtained in heat treatments B and D should be related to the absence of Ni_3Cb needle formation (Figure 8).

From the above analysis we conclude that significant improvement in crack growth resistance of Alloy 718 can be obtained by anneal at 926°C for 10 hours, furnace cool, and age at 734°C for 48 hours (heat treatment E). Improved crack growth resistance is obtained without sacrificing much in yield strength and tensile strength. It is possible that the above heat treatment could also provide good creep rupture strength comparable to the conventional heat treatment and this

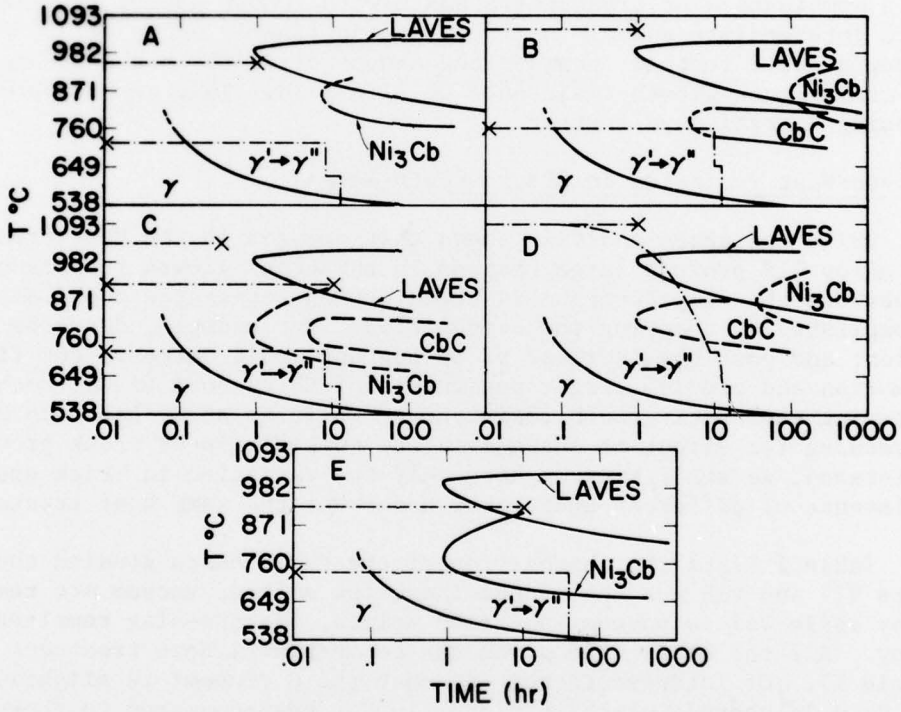

Fig. 8 - Transformation diagrams for Alloy 718 for different heat treatments.

is being investigated further. The high crack growth resistance of the above heat treatment appears to be related to the formation of large fractions of Ni_3Cb precipitates along the grain boundaries.

That large volume fraction of Ni_3Cb would also improve the high temperature notch sensitivity of Alloy 718 had been recognized by Muller and Donachie (9). In fact heat treatment C was devised to increase the Ni_3Cb content. Some of the Ni_3Cb formed during the intermediate anneal apparently dissolves during aging. This could be reprecipitated perhaps by aging for a long time like in heat treatment E. A combination of treatment C and E with 1038°C anneal (20 min), 926°C intermediate anneal for 10 hours and finally aging at 734°C for 48 hours could further increase the amount of Ni_3Cb that could improve the creep crack growth resistance of Alloy 718. This aspect, however, is being investigated further.

Heat-to-Heat Variation in Crack Growth Rates

While the above analysis shows that changes in the heat treatment for Alloy 718 produce large changes in the crack growth resistance, it is obvious that improvements in crack growth resistance could also be accomplished by changing the composition. For example, decrease of C content and possibly increase of Cb content would decrease CbC film formation and provide larger percentage of Cb to form Ni_3Cb, both improving the crack growth resistance. While no study has been made concerning the effect of changes in the composition on crack growth resistance, we shall present here only the variation in crack growth resistance of different heats that underwent the same heat treatment.

Table 2 lists the composition of the three heats studied thus far. Heats V17 and V18 are the vacuum induction melted, vacuum arc remelted alloy while V47 is vacuum induction melted, electro-slag remelted alloy. All the heats were given the conventional heat treatment A (Table 1). Of interest to note is that the C content is slightly lower and Cb + Ta content slightly higher in V47 heat compared to those in other two heats. The thickness of the specimens for the three heats are different. They are respectively 12.7, 15.34 and 25.4 mm. It has been shown recently, however, that the variation of specimen thickness from 2.6 to 25.4 mm has no effect on the crack growth rates in Alloy 718 under static and cyclic loads (15).

The crack growth rates versus stress intensity for the three heats are shown in Figure 9. The V18 data are the same as in Figure 2 while V17 data are from Ref. (3) and V47 data from Ref. (5). Within the scatter band the crack growth rates in V17 and V18 are the same and they are slightly higher than those in V47 heat. Better crack growth resistance of V47 could be due to either lower C and higher Cb + Ta content or to the difference in melting practice or due to both.

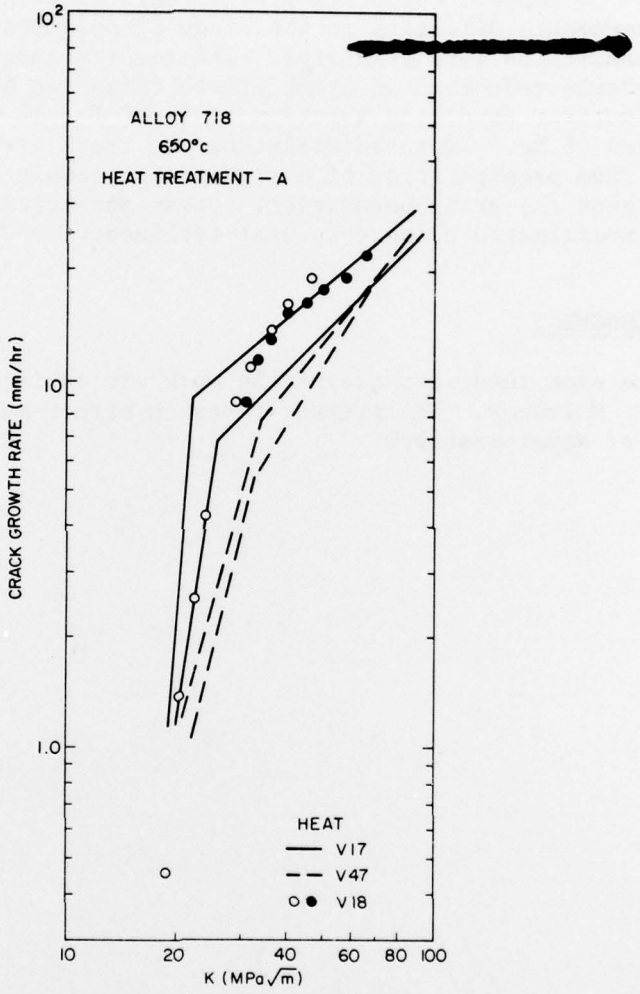

Fig. 9 - Comparison of crack growth rates for different heats.

SUMMARY AND CONCLUSIONS

Alloy 718 in its conventional heat treated form possesses very poor resistance to high temperature crack growth under static load. In order to improve the crack growth resistance in this alloy, five heat treatments differing in the range of solution temperature and aging conditions were evaluated. The results show that two orders of magnitude reduction in crack growth rates can be obtained by a heat treatment involving solution at 926°C for 10 hr and age at 734°C for 48 hr. Improved resistance to crack growth appears to result from precipitation of a significant amount of Ni_3Cb phase which strengthens the grain boundaries. Other mechanical properties are being investigated using this heat treatment.

ACKNOWLEDGMENT

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